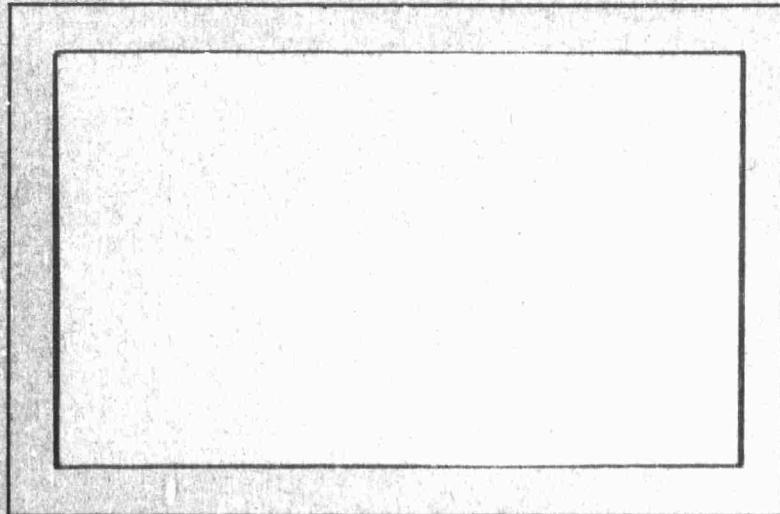
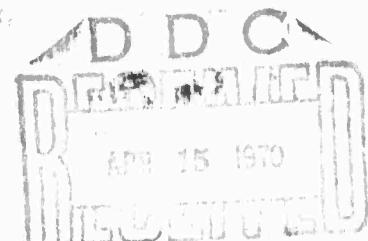


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**Stress-Corrosion Cracking Behavior
of an 18% Ni Maraging Steel**

by

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*** Now at Homer Research Laboratories, Bethlehem Steel Corp., Bethlehem, Pa.**
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ABSTRACT

Stress-corrosion cracking of an 18% Ni maraging steel in aqueous solutions was studied using precracked cantilever beam specimens. By appropriate heat treatments, six different structures having the same yield strength were obtained. Although significantly different plane strain fracture toughness values (K_{Ic}) resulted, it was found that the threshold plane strain stress intensity (K_{Iscc}) was the same for all structures. K_{Iscc} had the same value in 3% NaCl at various pH values, in 1N H_2SO_4 , and in distilled water. Specimens tested in 3% NaCl under both anodic and cathodic applied potentials also exhibited this same K_{Iscc} value. Fractographic inspection of the crack surfaces revealed no apparent differences due to changes in solution, pH, or applied potential. The crack path was intergranular in all cases. However, specimens austenitized at 1500°F exhibited crack branching, whereas in specimens austenitized at much higher temperatures branching no longer occurred. Aging time and temperature seemed to change only the time to failure. The mechanism most consistent with all observations appears to be hydrogen cracking.

INTRODUCTION

The stress-corrosion cracking behavior of high-strength maraging steel has not been investigated extensively. Failure of maraging or other high-strength martensitic steels in aqueous solutions does not seem to depend on the presence of a specific ion, as is the case, for example, in the austenitic stainless steels. The only safe generalization that can be made to date about the cracking behavior of maraging steels is that higher yield strength tends to increase the susceptibility to cracking.

Many of the previous investigations into the stress-corrosion behavior of maraging steels are summarized in the reviews by Matteoli and Sonya¹, Phelps², and Kennedy and Whittaker³. Even though the heat treatment of maraging steel affects its electrochemical characteristics, almost all these previous investigations were concerned with determining the effects of either electrochemical variables or metallurgical variables but not both.

The investigation described below was therefore designed to study the effects of both types of variables simultaneously. By working with a number of different microstructures having identical yield strength, it was possible to consider the effects produced on cracking behavior through variations of austenite grain size, amount of austenite, precipitate size and distribution, and fracture toughness. Electrochemical variables were studied simultaneously by varying the test solution, pH, applied potential, and atmosphere above the solution.

Since the initiation of a crack usually occurs at a rate many orders of magnitude slower than the propagation rate, any test involving smooth specimens is essentially a measure of the resistance of the material to crack initiation, e.g., by pitting. To avoid the long and poorly reproducible initiation period of the stress-corrosion process, a sharp artificial crack was incorporated into the test specimen.

Another important feature of precracked tests is that the concepts of linear elastic fracture mechanics can be applied to describe mathematically the effects of geometry and stress on conditions at the crack tip. Brown⁴ has recently reviewed the application of fracture mechanics to stress-corrosion testing; all symbols used in this paper will have the same definitions as presented in Brown's review.

EXPERIMENTAL PROCEDURES

The steel investigated was an 18% Ni (300 grade) maraging type purchased in the form of hot-rolled and straightened bar. All experiments were performed using material from a single heat. The composition, as determined by the manufacturer is given in Table I.

Heat treating was done in neutral salt baths. The two austenitizing treatments examined were (a) an anneal of 2 hours at 1500°F and (B) an anneal of 1 hour at 2300°F followed by direct quenching to 1700°F and holding for 4 hours. Three aging treatments, each of which hardened the previously austenitized steel to Rc 52-53 were selected: (1) 800°F for 10 hours, (2) 900°F for 3 1/2 hours, and (3) 900°F for 100 hours.

The stress-corrosion test specimens were cantilever beams approximately 1/2" x 5/8" in cross section and 12" long. A V-notch 0.22" deep with a root radius of less than 0.01" was machined 5 1/2" from one end. The specimens were also side-grooved on each side to a depth of 5% of the width (approximately 0.025"). After heat treatment the specimens were precracked on a constant load fatigue machine. The procedures outlined above for obtaining precracked cantilever beam specimens all conform to published recommendations for insuring plane strain conditions at the crack tip⁵.

Testing was done in (i) distilled water, (ii) 3% NaCl of various pH values, or (iii) 1N H₂SO₄. Reagent grade chemicals and distilled water were used to prepare the salt and acid solutions. Each solution was placed in a sealed one-half pint polyethylene container surrounding the crack. Provision in these cells was made for a gas inlet and outlet tube and occasionally for a Luggin probe and an auxiliary platinum electrode.

The procedure for testing the precracked specimens involved: (1) adding solution to the cell, (2) bubbling hydrogen or oxygen gas into the solution for 15-60 min. (3) bolting a 30 in. lever arm onto the free end of the specimen, and (4) adding the desired load to the end of the lever arm. In controlled potential tests, the potential was applied potentiostatically at least 10 min. before the load was applied.

An approximate indication of crack growth was obtained by monitoring the deflection of the cantilever with a dial gage. Time to failure was recorded when the falling lever arm tripped a microswitch.

Data obtained from these precracked cantilever tests are plotted as initial stress intensity, K_{Ii} , versus the time to failure, t_f . K_{Ii} can be calculated from any of several expressions which Rolfe et al.⁶ have shown can be simplified to the form illustrated in Figure 1. The various treatments differ slightly in the value given the $f(\frac{a}{w})$ term.

To establish the threshold stress intensity, K_{Iscc} , the step loading technique was employed. If no crack growth was evident after 10,000 minutes the load was increased slightly. It has previously been shown that this technique does not alter the K_{Iscc} value from that obtained by direct loading⁷.

The toughness of the steel was also determined from cantilever beams. Several specimens from each heat treatment were loaded to failure in air. The stress intensity at fracture, K_{Ix} , gives an approximation of the plane strain stress intensity (a measure of fracture toughness), K_{Ic} , which is normally determined by more elaborate methods.

After a stress corrosion failure the fractured surfaces of many specimens were inspected directly with a scanning electron microscope. In some instances when the surfaces were less rough, plastic-carbon replicas were examined in a transmission electron microscope.

RESULTS

Microstructural Variations

Microstructures produced by the six different heat treatments were documented by standard techniques. The prior austenite grain size was determined by comparing optical micrographs with standard ASTM grids. Austenitization at 1500°F resulted in a grain size of ASTM #9 while austenitizing at 2300°F + 1700°F yielded a grain size of ASTM #0. A substantial variation in size between individual grains was noted within each specimen, especially after the high temperature anneal.

The fine structure resulting from the heat treatments was observed by thin film electron microscopy. Both austenitizing treatments produced a typical martensitic structure composed of laths with a high density of dislocations⁸.

Aging at 800°F for 10 hr. did not change the appearance of this structure to any extent. The dislocation density seemed slightly lower, but no precipitates were visible at 60,000X (Kinetic studies by other investigators indicate the presence of two types of Mo-precipitate below 850°F⁹).

Aging 3 hrs. at 900°F resulted in two types of precipitate in the martensite matrix. These precipitates have been previously identified as Ni₃Mo and σ - FeTi⁹. Aging for 100 hrs. at 900°F results in larger precipitates and approximately 20% reverted austenite.

When the higher temperature austenization treatment was employed large grain boundary precipitates were observed in addition to the usual matrix precipitates. These large precipitates are believed to be Ti(C, N)¹⁰.

Fracture Toughness Variations

Table II presents the K_{Ic} values obtained for each of the six heat treatments selected for investigation. It is evident that austenitizing at higher temperatures for longer times results in a more brittle structure. Aging at 800°F results in better toughness than aging at 900°F, and overaging gives a lower toughness relative to underaging. The toughness variations, however, are not related in the stress-corrosion cracking behavior.

Cracking Behavior

Sodium Chloride Solutions

Figure 2 illustrates the cracking behavior in deaerated 3% NaCl at pH 6.3 of maraging steel annealed at 1500°F and aged. The time to failure appears relatively insensitive to stress intensity but very dependent on aging treatment. K_{Iscc} has a value of approximately 10-15 Ksiv/in. for all three aging treatments. When the pH of the salt solution is changed to 1.7 the same type of behavior is observed (Figure 3). K_{Iscc} is still 10-15 Ksiv/in.

Figure 4 and Figure 5 illustrate the cracking behavior in deaerated 3% NaCl of pH 6.3 and pH 1.7 respectively of maraging steel annealed at 2300°F + 1700°F and aged. Time to failure appears much more dependent on stress intensity and appreciably less so on aging treatment. K_{Iscc} , however, still has a value of 10-15 Ksiv/in.

The mode of cracking in all cases was intergranular with respect to the prior austenite grains. Steel annealed at 1500°F exhibited crack branching whenever the stress intensity at the crack tip exceeded approximately 31 Ksiv/in., whereas annealing at the higher temperatures never resulted in branching.

Branching was in the form of two divergent cracks, approximately 45° (measured at the side surface) to the plane of the fatigue crack. These two cracks continued to branch further until the structure shown in Figure 6 resulted.

Sulfuric Acid Solutions

Several experiments were conducted in H_2 -saturated 1N H_2SO_4 . The stress

intensity-failure time plots were similar to those obtained in 3% NaCl for material given equivalent heat treatments (Figure 1, 7). Branching was also observed in specimens given the 1500°F annealing treatment.

Distilled Water

Deaerated distilled water was also used as a test environment. The resulting stress intensity-failure time curves were similar to those obtained in the salt and acid solutions. Figure 8 presents a comparison of cracking behaviors in distilled water, sulfuric acid, and sodium chloride for a single heat treatment. The stress intensity dependence and the threshold stress intensity appear unaffected by environment.

Electrode Potential Variations

Most previous investigations on stress-corrosion cracking of martensitic steels have relied on electrochemical concepts to suggest the controlling mechanism. The effect of potential on cracking behavior in 3% NaCl of pH 6.3 for one heat treatment is shown in Figure 9. All applied potentials lower t_f with respect to the unpolarized condition at all K_{Ii} . Thus the electrochemical means of distinguishing between possible mechanisms¹¹ does not yield valid results. K_{Iscc} , however, does not appear to be measurably affected by applied potential.

Further Tests

Since the threshold stress intensity for this maraging steel appears invariant on the basis of the above results, further tests were conducted to determine K_{Iscc} in other solutions; (a) in 3% NaCl at other pH values, (b) in oxygen-

saturated 3% NaCl, and (c) in 3% NaCl + 1.5% $\text{Na}_2\text{Cr}_2\text{O}_7$ at pH 6.1. Figure 10 presents the threshold stress intensities obtained for all tests conducted. The value is always about 10-15 $\text{Ksi}/\text{in.}$

Fractography

Inspection by scanning electron microscopy of the surfaces of fractured stress-corrosion specimens annealed at 1500°F revealed that in all cases failure is intergranular, the individual grain facets being rather featureless. Many secondary cracks or crevices were always evident. The surface appearance did not seem to depend on aging treatment, test solution, pH, or applied potential and always appeared as in Figure 11.

Austenitizing at the higher temperatures (2300°F + 1700°F) did change the surface appearance, however, as illustrated in Figure 12. Cracking was still intergranular with many secondary cracks present, but the individual grain facets were no longer smooth. They appeared to have many tear ridges, similar to those attributed to quasi-cleavage¹².

The severe roughness of the stress-corrosion fracture surfaces made preparation of replicas very difficult, especially for material annealed at the higher temperatures. Replicas obtained from material annealed at 1500°F showed features characteristic of stress-corrosion fractures in high strength martensitic steels (Figure 13). Although a thorough examination of all specimens was not conducted, Figure 13 is typical of all tests run on 1500°F annealed specimens.

DISCUSSION

The main experimental observations from this study are:

1. K_{Iscc} was found to be constant throughout all tests and has a value of 10 to 15 ksi/in.
2. Low temperature aging (800°F) results in shorter failure times than high temperature aging (900°F). At 900°F underaging (3 1/2 hrs.) is more beneficial than overaging (100 hrs.).
3. Cracking occurs intergranularly in all cases. Austenitizing at 1500°F resulted in branching whenever K_I exceeded approximately 31 ksi/in. Material austenitized at higher temperatures (2300°F + 1700°F) was never observed to exhibit crack branching.
4. Electrochemical and fractographic techniques did not provide a useful means for determining what mechanism or mechanisms may control the stress-corrosion process.

The two mechanisms usually proposed to explain stress-corrosion cracking in martensitic steels are (1) an anodic dissolution mechanism or (2) a hydrogen cracking mechanism. The experimental observations appear to be most easily explained by the latter mechanism.

Since the theories of hydrogen cracking depend on the amount of hydrogen present in critical areas and not on its source, the exact nature of the environment should have little influence as long as the necessary amount of hydrogen can enter the steel. The similar stress intensity-failure time curves observed in sulfuric acid, sodium chloride, and distilled water tend to confirm this statement if hydrogen is actually available in each of these environments.

Immersion of maraging steel into sulfuric acid results in the evolution of hydrogen on the steel surface and hence probable entry of some hydrogen into the matrix (H_2SO_4 with arsenic additives is a solution often used to cathodically charge hydrogen into metal).

Recent evidence also indicates that hydrogen can enter the metal at the crack tip in sodium chloride solutions. Brown¹³ has determined the solution chemistry near the tip of a propagating crack in several steels (including a maraging steel) tested in sodium chloride at various initial pH values. In all cases the pH at the crack tip was 3.6-3.8 and the ferrous ion concentration about 1 molar. If similar behavior in distilled water is assumed, then inspection of the Pourbaix diagram for the iron-water system indicates that the thermodynamic requirements for the generation of hydrogen are met within the crack.

The effect of aging treatment on failure time must be accounted for by assuming that the resulting structures affect either the hydrogen entry rate or the hydrogen diffusion rate. If the tests of material annealed at 1500°F are considered it is seen that time to failure is changed more by varying aging treatment than by varying test solution or applied potential. Since these latter two variables cannot change the diffusion rate of hydrogen in the matrix it appears that aging influences the time to failure by affecting the hydrogen diffusion rate. This situation is consistent with the known fact that structural changes affect the trapping of hydrogen in maraging steels¹⁴.

Usually crack branching is associated with the fracture of brittle materials. The theory of branching is based upon a redistribution of stresses at the tip of a high velocity crack in which the maximum normal stresses occur on two planes inclined to the plane of the running crack.¹⁵ Mechanistically this requires the spreading of an "advance" crack already situated on these planes. Further analysis indicated that a constant crack velocity¹⁶ or a critical

stress intensity¹⁷ must be achieved for branching to occur.

Recently stress-corrosion crack branching has been reviewed by Carter¹⁸. The requirements for branching were found to be similar to those just described. Whenever a steel exhibited branching the range of constant crack velocity extended from approximately K_{Isc} to K_{Ic} . The critical stress intensity for branching, K_{Ib} , was found to be two to four times the threshold stress intensity. This implies that when the ratio of K_{Isc}/K_{Ic} is too large, terminal fracture will occur before K_{Ib} is reached. If a mechanism similar to that postulated for brittle branching is also assumed, then a hydrogen mechanism is the most probable method by which the "advance" crack can be spread.

There are objections to using the "advance" crack mechanism to explain stress-corrosion crack branching. Although dial gage observations indicated that crack growth may have occurred in steps of high velocity there were no surface markings indicating crack arrests and the average velocities were very slow (10^{-3} - 10^{-5} in/min.). It is unlikely then that any velocity induced stress distribution takes place. Interestingly enough, the more brittle specimens (those given the high temperature anneal) exhibited higher crack velocities but no branching. In those cases where $K_{Ib} < K_{Ic}$, the velocity did not achieve a constant value until after some growth, which indicates that a constant velocity may be a result of branching rather than a contributing cause.

Carter's suggestion¹⁸ that K_{Ib} is two to four times K_{Isc} is approximately satisfied for the fine grained material, but quite probably not for that with a coarse austenite grain size. (The full four times a K_{Isc} at the high end

of the observed range would be necessary for $K_{Ib} > K_{Ic}$). Since K_{Isc} and K_{Ic} are both structure-sensitive properties, the explanation for branching may also be sought in structural terms, in particular, structural differences resulting from the various annealing treatments. This approach receives support from the existence of a branching threshold (K_{Ib}) which suggests a possible interaction between a characteristic size of the structure and the plastic zone size.

Calculation of the plastic zone diameter gives the following¹⁹:

$$p \approx \frac{1}{3\pi} \left(\frac{K_{Ib}}{\sigma_{ys}} \right)^2 = \frac{1}{3\pi} \left(\frac{31}{270} \right)^2 = 1.4 \times 10^{-3} \text{ in.}$$

This value is close to the grain diameter of the material (ASTM #9 = 0.625×10^{-3} in.), certainly within the accuracy of K_{Ib} measurements (range 21-40; avg. 31). In material annealed at the higher temperatures in which no branching was observed, a K_{Ib} of 98 ksi/in. would be needed to achieve a plastic zone diameter equal to the grain size (ASTM #0 = 14.1×10^{-3} in.). Since the highest K_{Ic} measured in this material was less than 60 ksi/in. fracture would occur before branching. Thus the experimental observations are consistent with branching being determined by interaction between the plastic zone at the crack tip and some characteristic structure size which is related to or approximated by the prior austenite grain size.

The electrochemical means of distinguishing between anodic dissolution stress-corrosion and hydrogen cracking depends on the imposition of applied potentials. If the hydrogen generated by the cathodic reaction of a corrosion process is the cause of cracking then the application of a cathodic potential should increase the amount of hydrogen and shorten time to failure. Conversely,

application of an anodic potential should decrease the available hydrogen and lengthen time to failure. If anodic dissolution is responsible for cracking the application of a cathodic potential should lengthen time to failure while an applied anodic potential should shorten time to failure (if attack remains localized). The various types of curves which may be encountered in use of the polarization method have been summarized by Bhatt and Phelps¹¹.

The experimentally observed decrease in failure time with both anodic and cathodic applied potentials is interpreted as indicating an anodic dissolution mechanism at anodic potentials and a hydrogen cracking mechanism at cathodic potentials. No direct information on the operating mechanism under open-circuit conditions is obtained in this type of experiment.

A change of mechanism with potential does not appear consistent when K_{Isc} is independent of potential. Recent literature has suggested that hydrogen may be generated even on specimens anodically polarized^{20, 21}. If this occurred, K_{Isc} should remain unaffected by potential even though the time to failure decreases with both anodic and cathodic applied potentials.

The assumption of hydrogen generation at anodic potentials may not be necessary if we consider that the potential at the crack tip may be quite different from the controlled bulk potential. The width of the crack can be estimated from fracture mechanics as $\delta = 4K_{Ib}^2 / \pi E \sigma_{ys} = 0.0002$ inch, assuming that the stress intensity at branching is never exceeded. The extreme narrowness of this crack suggests, in accordance with theoretical treatments^{23, 24}, that much of the crack tip is relatively unaffected by the value of the potential outside the crack.

If the crack tip potential is unchanged then the solution chemistry in this region will always be suitable for hydrogen entry into the metal. A hydrogen mechanism could then operate regardless of the applied bulk potential and K_{Iscc} should remain independent of potential.

The shortened time to failure observed when a potential was applied is probably a reflection of the fact that a portion of the hydrogen responsible for cracking may have entered the steel in regions easily influenced by the applied potential (e. g. the crack edges or the upper crack surfaces). Since the diffusion coefficient of hydrogen is high (10^{-6} - 10^{-7} in 2 /sec.) in martensite, transport of hydrogen to the crack tip will not be limiting. The exact failure time-applied potential relationship will be determined by bulk solution properties such as the equilibrium hydrogen discharge potential, whether the corrosion potential is noble or active with respect to the discharge potential, and if hydrogen can be generated under anodic potentials. Any polarization which induces hydrogen generation will be detrimental, regardless of whether it is anodic or cathodic.

Additional support for a single mechanism causing cracking is provided by the similarity of the fracture surfaces under all conditions.

The stress-corrosion process in this steel can be hypothesized as follows: Application of a load to the specimen opens the crack and allows solution to be drawn to the crack tip. The enhanced dissolution of metal from the crack tip changes the character of the nearby solution, probably by hydrolysis of the various dissolving metallic ions, $M^{+x} + yH_2O \rightarrow M(OH)_y^{x-y} + yH^+$. The acidic nature of this crack tip solution means that the predominant cathode reaction will be hydrogen reduction. Thus a source of hydrogen is always available for causing hydrogen cracking.

Due to the high diffusion coefficient of hydrogen in martensite a significant portion of the total hydrogen used in cracking can originate away from the crack tip region; thus polarization affecting this source can alter cracking kinetics to some degree, but cannot stop stress-corrosion.

SUMMARY

This investigation has shown that the heat treatment given a maraging steel, through its effect on the resulting structure, can have a pronounced influence on stress-corrosion cracking behavior. Annealing and aging treatments were varied in a manner designed to keep the yield strength constant. The threshold stress intensity was found to be independent of heat treatment, test solution, pH value, and applied potential. Such behavior suggests that a single mechanism-hydrogen cracking-is responsible for stress-corrosion.

Although the threshold stress intensity was invariant, the time to failure could be changed slightly by varying the test solution, to a somewhat greater extent by varying the pH or applied potential, and by two orders of magnitude by varying the aging treatment. The hydrogen cracking mechanism was shown to be consistent with these failure time changes. Aging treatments were very effective in affecting failure times because the diffusion coefficient of hydrogen in martensite can be changed by the number and type of trapping sites which are dependent on aging treatment.

Fractographic analysis also indicated that a single mechanism probably causes cracking in all cases, since the fracture surfaces did not change in appearance as the test conditions were varied.

Another influence of structure on stress-corrosion behavior is illustrated by the phenomenon of crack branching. Branching was found to occur when the crack-tip plastic zone interacted with some structural unit which appears to be related to the prior austenite grain size. Fine-grained material which is produced by lower austenitizing temperatures branches extensively thus increasing the time to failure relative to the non-branching case.

Table I. COMPOSITION OF MARAGING ALLOY

Element	C	Mn	P	S	Ni	Co	Mo	Ti	Al	Si	Zr	Fe
Wt. %	0.005	0.01	0.004	0.006	18.23	8.99	5.08	0.67	0.08	0.01	0.027	Bal.

Table II. EFFECT OF HEAT TREATMENT ON K_{Ix}

<u>Heat Treatment</u>	<u>K_{Ix} (ksi/in.)</u>	<u>Average</u>
1500 F/2 hr. + 800 F/10 hr.	122, 115.5 122	119.9
1500 F/2 hr. + 900 F/3 1/2 hr.	102.5, 92, 112, 90.5	99.2
1500 F/2 hr. + 900 F/100 hr.	68, 71.5	69.8
2300 F/1 hr. + 1700 F/4 hr. + 800 F/10 hr.	58.5, 61.5, 55.6	58.5
2300 F/1 hr. + 1700 F/4 Hr. + 900 F/3 1/2 hr.	53.5, 59.5 60.0	57.5
2300 F/1 hr. + 1700 F/4 hr. + 900 F/100 hr.	52.3, 51.5, 54.2	52.6

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FIGURE CAPTIONS

Fig. 1. Cantilever Beam Specimen

Fig. 2. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 6.3 of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F.

Fig. 3. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 1.7 of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F.

Fig. 4. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 6.3 of 18% Ni Maraging Steel Austenitized 1 hr. at 2300° F + 4 hrs. at 1700° F.

Fig. 5. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 1.7 of 18% Ni Maraging Steel Austenitized 1 hr. at 2300° F.

Fig. 6. Extended Crack Branching

Fig. 7. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 1N H₂SO₄ of 18% Ni Maraging Steel Austenitized 2 hr. at 1500° F.

Fig. 8. Effect of Environment of Stress-Corrosion Susceptibility of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F and Aged 10 hrs. at 800° F.

Fig. 9. Effect of Applied Potential on Stress-Corrosion Susceptibility of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F and Aged 3 1/2 hrs. at 900° F.

Fig. 10. K_{Iscc} of 18% Ni Maraging Steel in Various Heat Treated Conditions and Environments

Fig. 11. Typical Stress-Corrosion Topography of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F (SEM 1800X).

Fig. 12. Typical Stress-Corrosion Topography of 18% Ni Maraging Steel Austenitized 1 hr. at 2300° F + 4 hrs. at 1700° F (SEM 1800X).

Fig. 13. Typical Stress-Corrosion Electron Fractograph of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F (3250X).

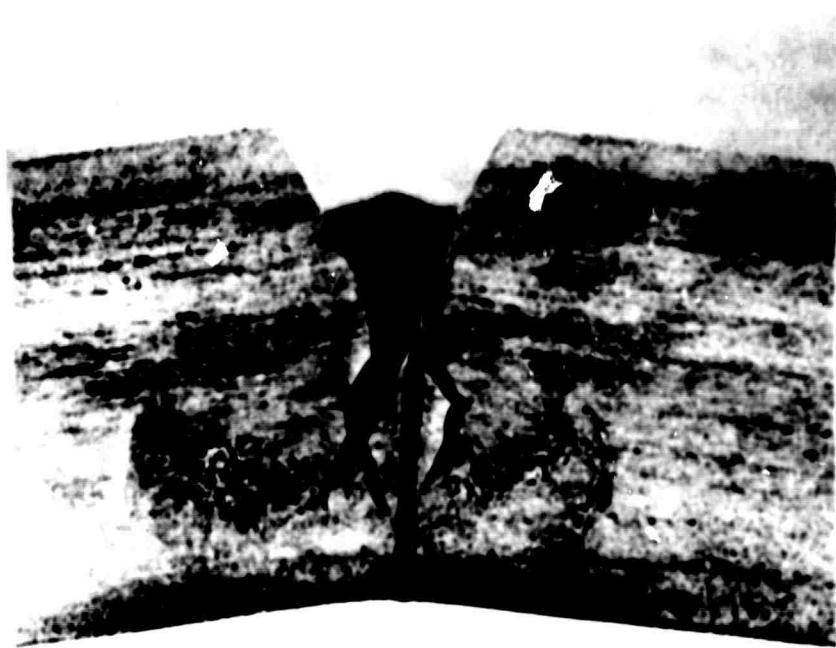


Fig. 6. Extended Crack Branching

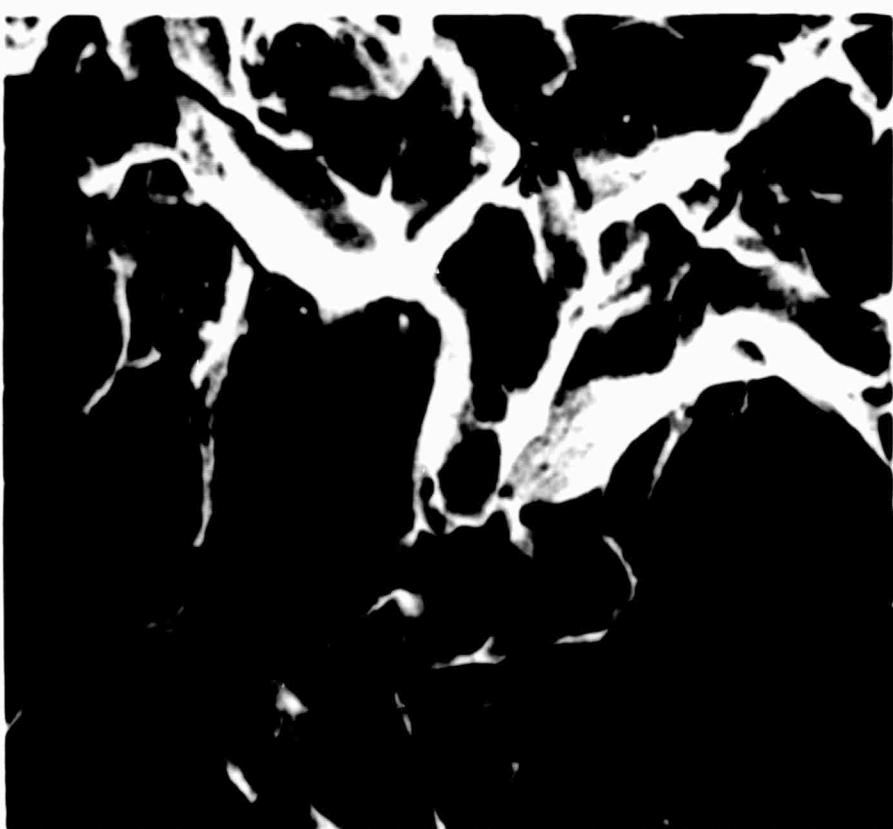


Fig. 11. Typical Stress-Corrosion Topography of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500°F (SEM 1800X).

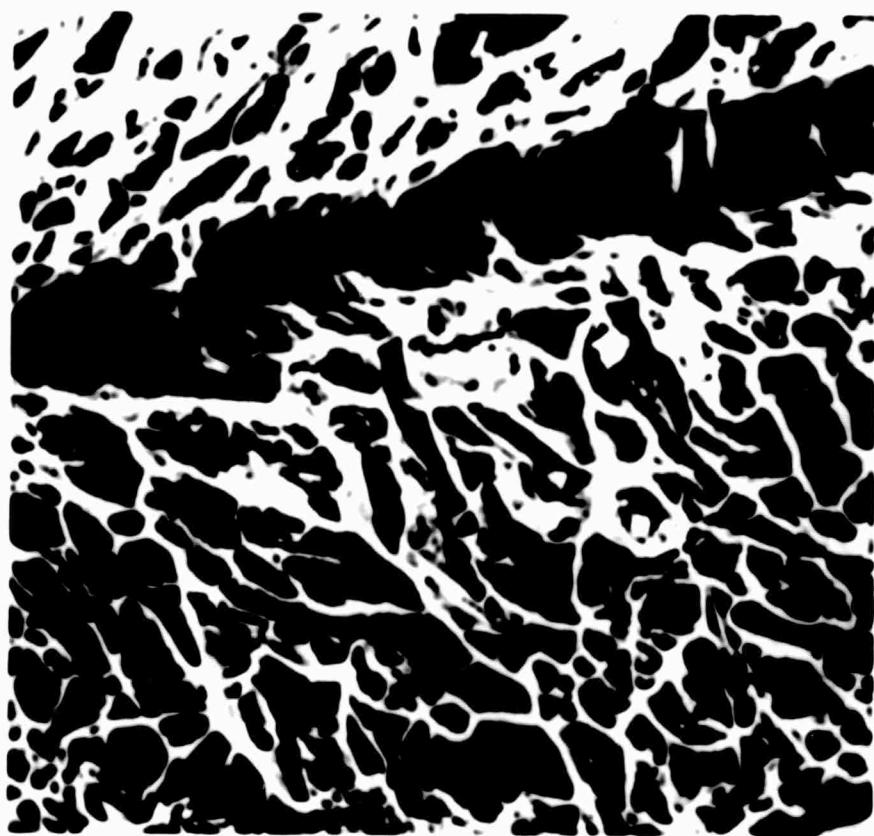
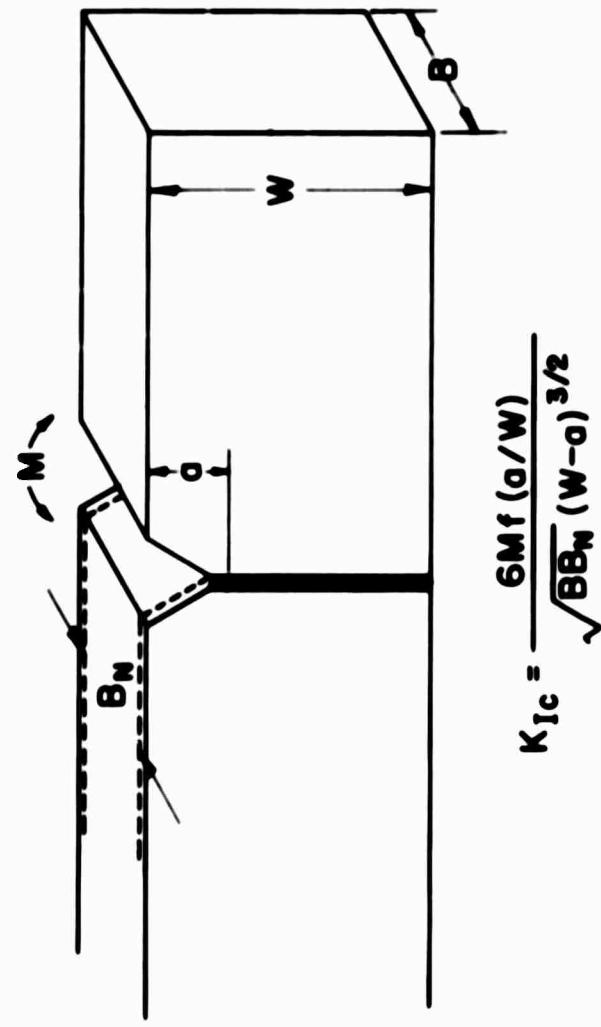


Fig. 12. Typical Stress-Corrosion Topography of 18%
Ni Maraging Steel Austenitized 1 hr. at
2300°F + 4 hrs. at 1700°F (SEM 1800X).



Fig. 13. Typical Stress-Corrosion Electron Fractograph of 18 Ni Maraging Steel Austenitized 2 hrs. at 1500° F (3250X).



$$K_{Ic} = \frac{6Mf(a/W)}{\sqrt{BB_N}(W-a)^{3/2}}$$

Fig. 1. Cantilever Beam Specimen

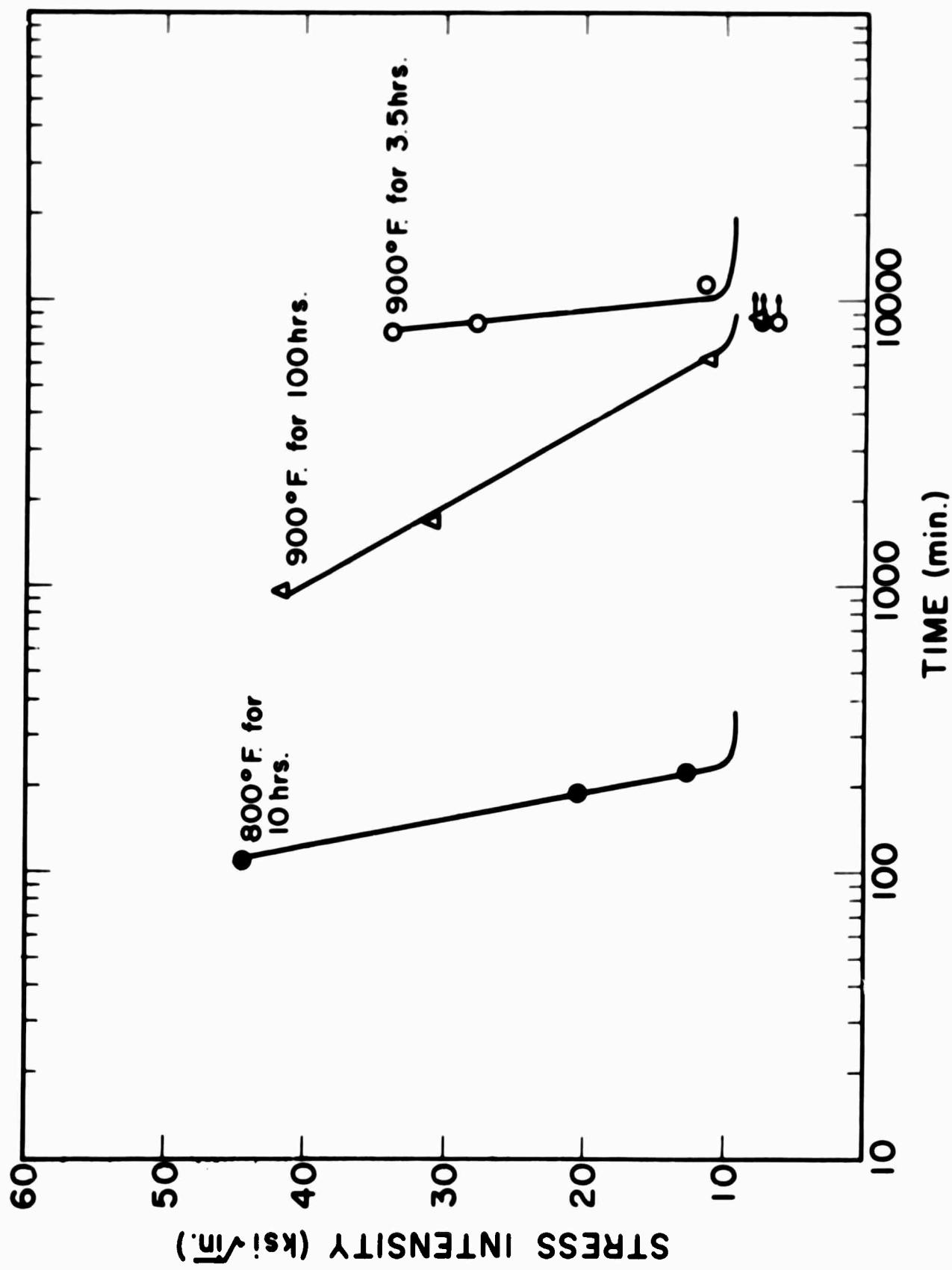


Fig. 2. Effect of Aging Treatment on Stress-Corrosion Susceptibility
in 3% NaCl at pH 6.3 of 18% Ni Maraging Steel Austenitized
2 hrs. at 1500°F.

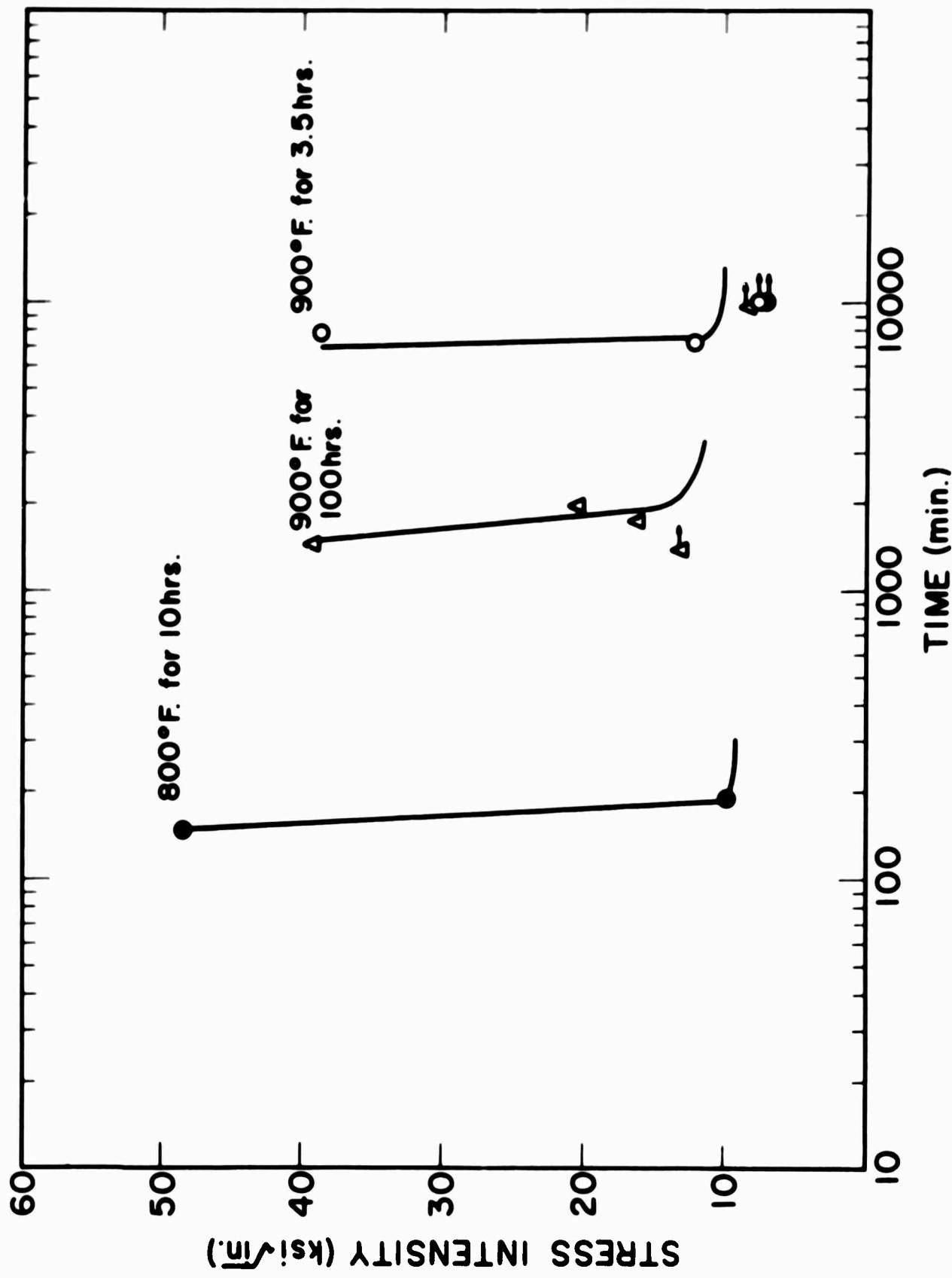


Fig. 3. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 1.7 of 18 Ni Maraging Steel Austenitized 2 hrs. at 1500°F.

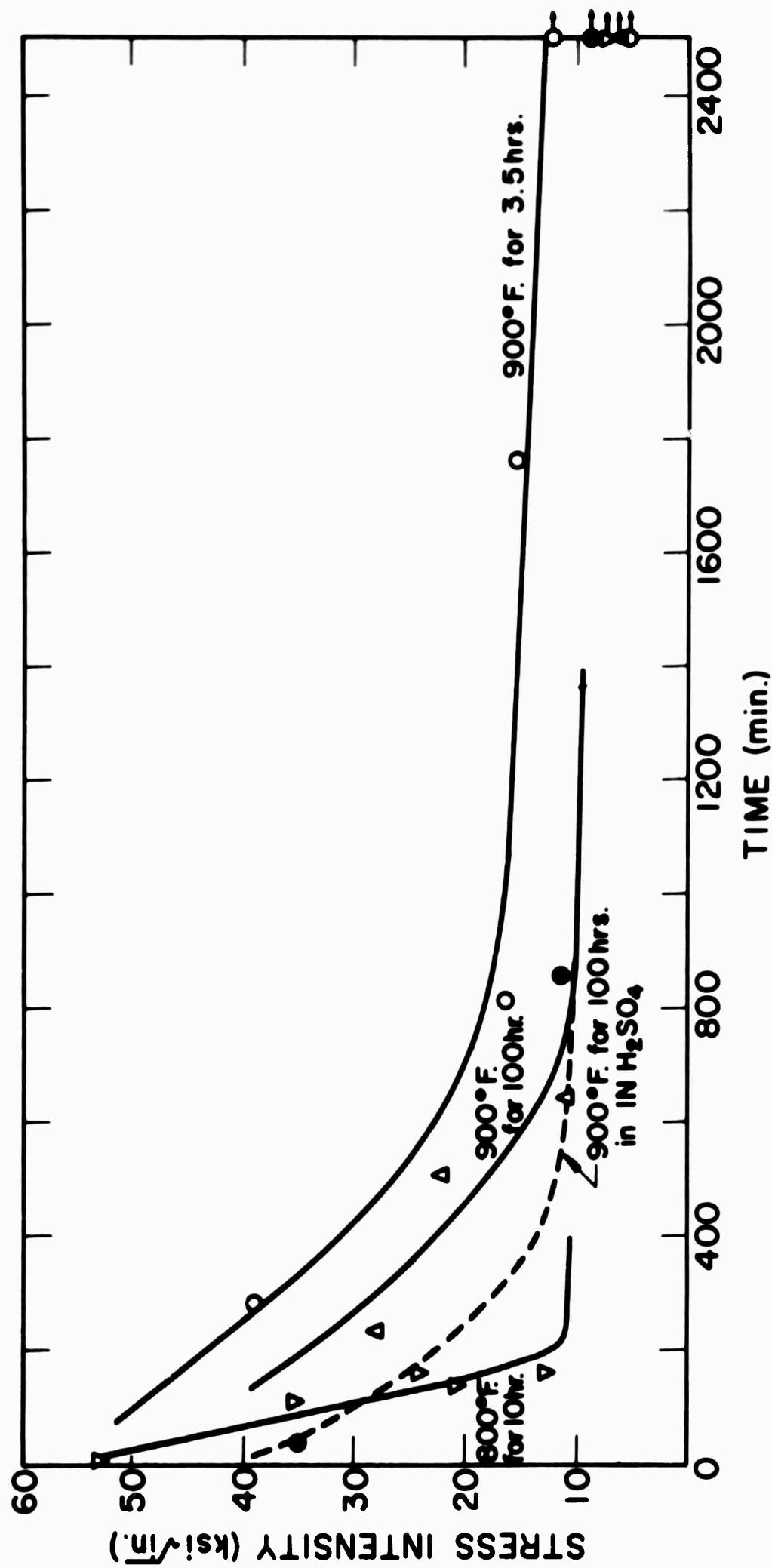


Fig. 4. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 6.3 of 18% Ni Maraging Steel Austenitized 1 hr. at 2300°F + 4 hrs. at 1700°F.

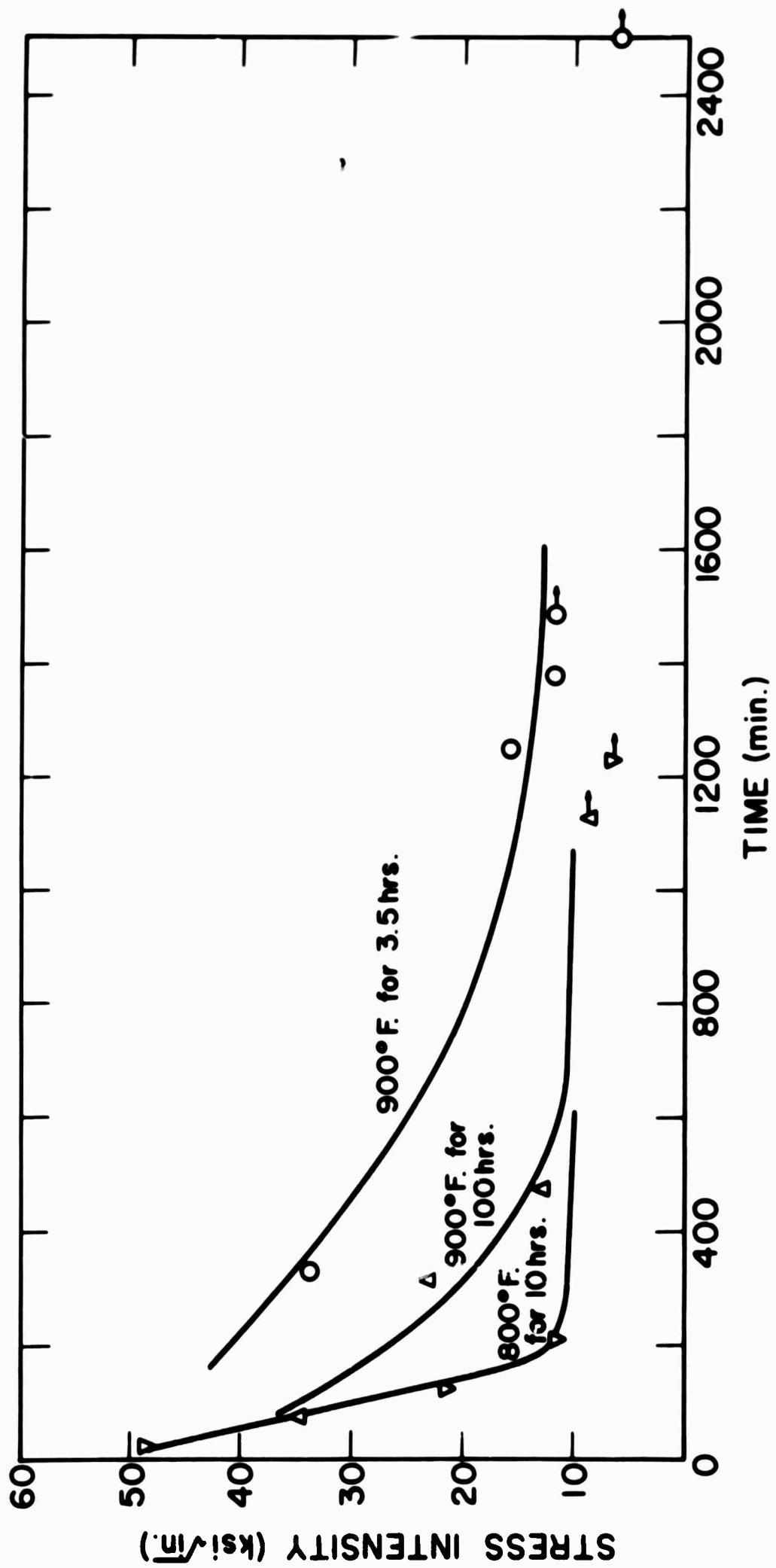


Fig. 5. Effect of Aging Treatment on Stress-Corrosion Susceptibility in 3% NaCl at pH 1.7 of 18% Ni Maraging Steel Austenitized 1 hr. at 2300°F.

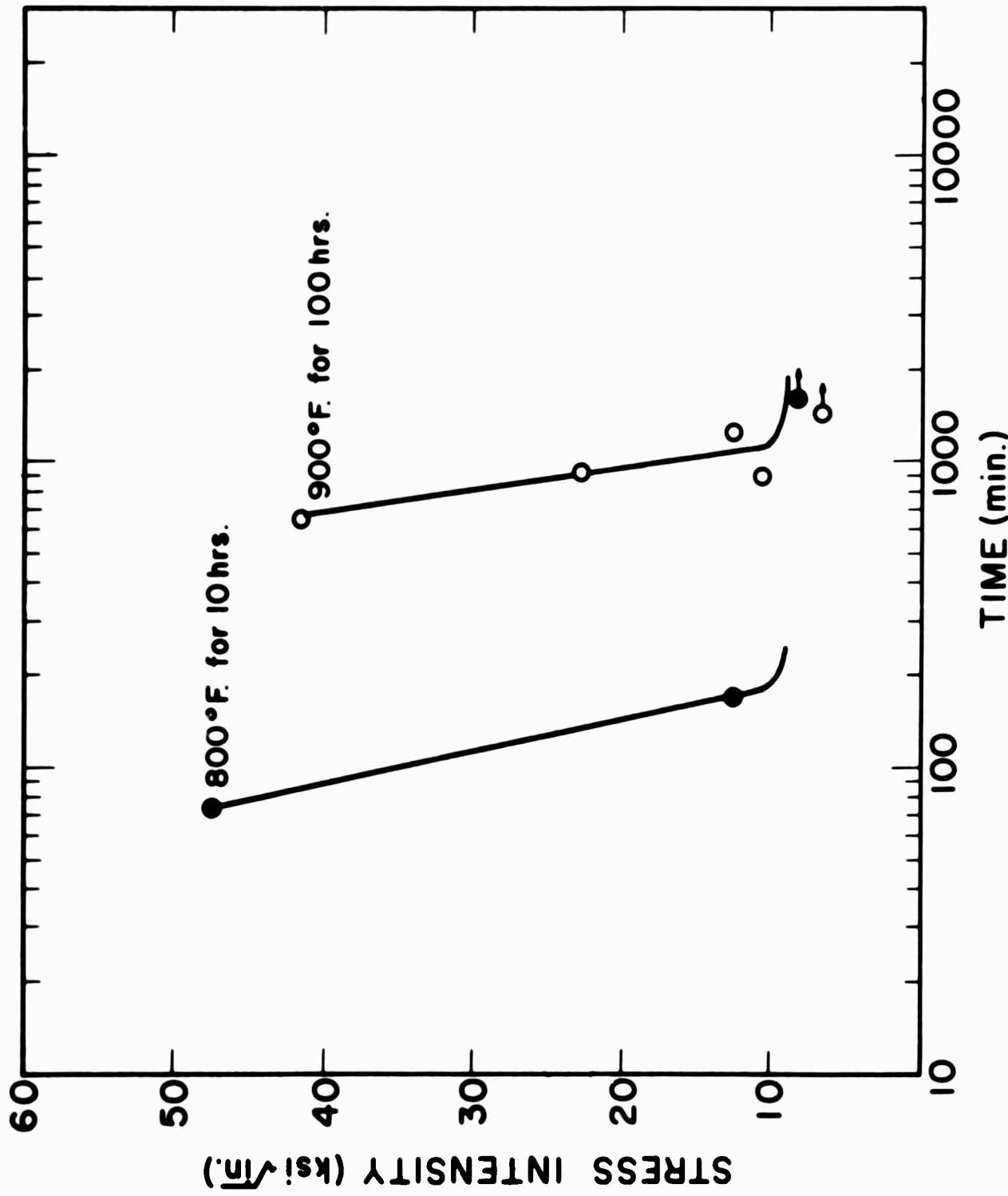


Fig. 7. Effect of Aging Treatment on Stress-Corrosion Susceptibility in
IN H_2SO_4 of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500° F.

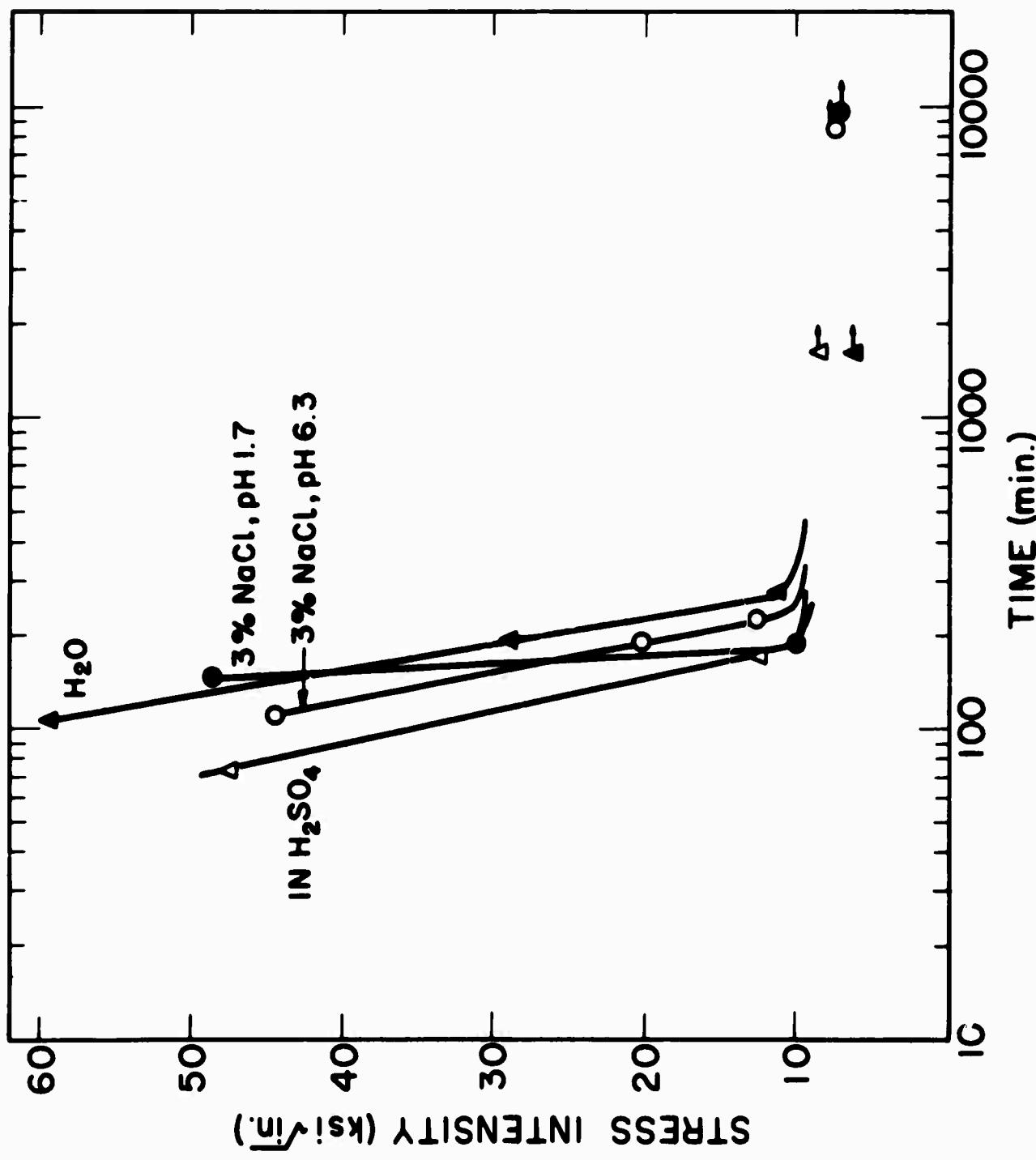


Fig. 8. Effect of Environment on Stress-Corrosion Susceptibility of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500°F and Aged 10 hrs. at 800°F.

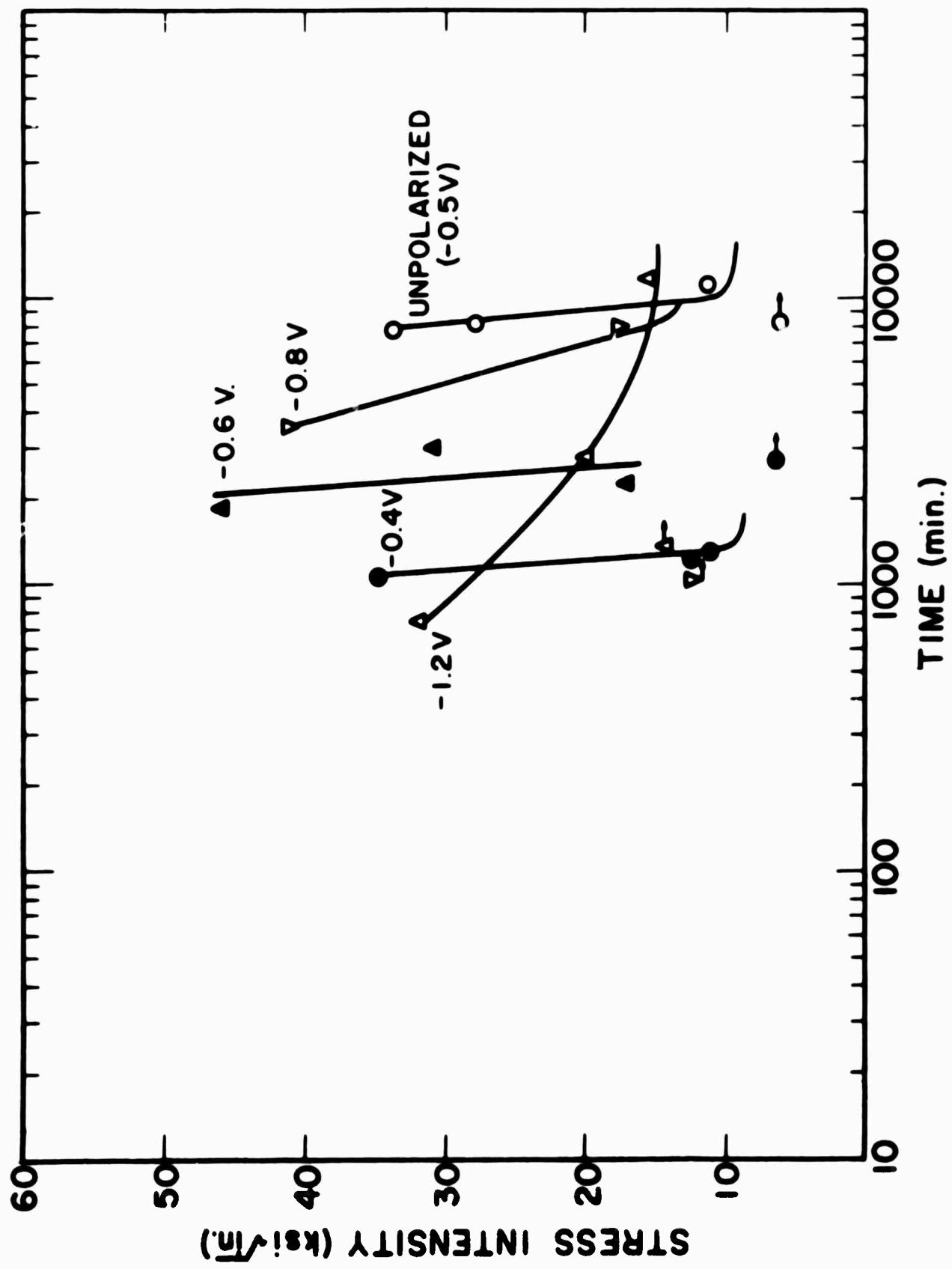


Fig. 9. Effect of Applied Potential on Stress-Corrosion Susceptibility of 18% Ni Maraging Steel Austenitized 2 hrs. at 1500°F and Aged 3 1/2 hrs. at 900°F.

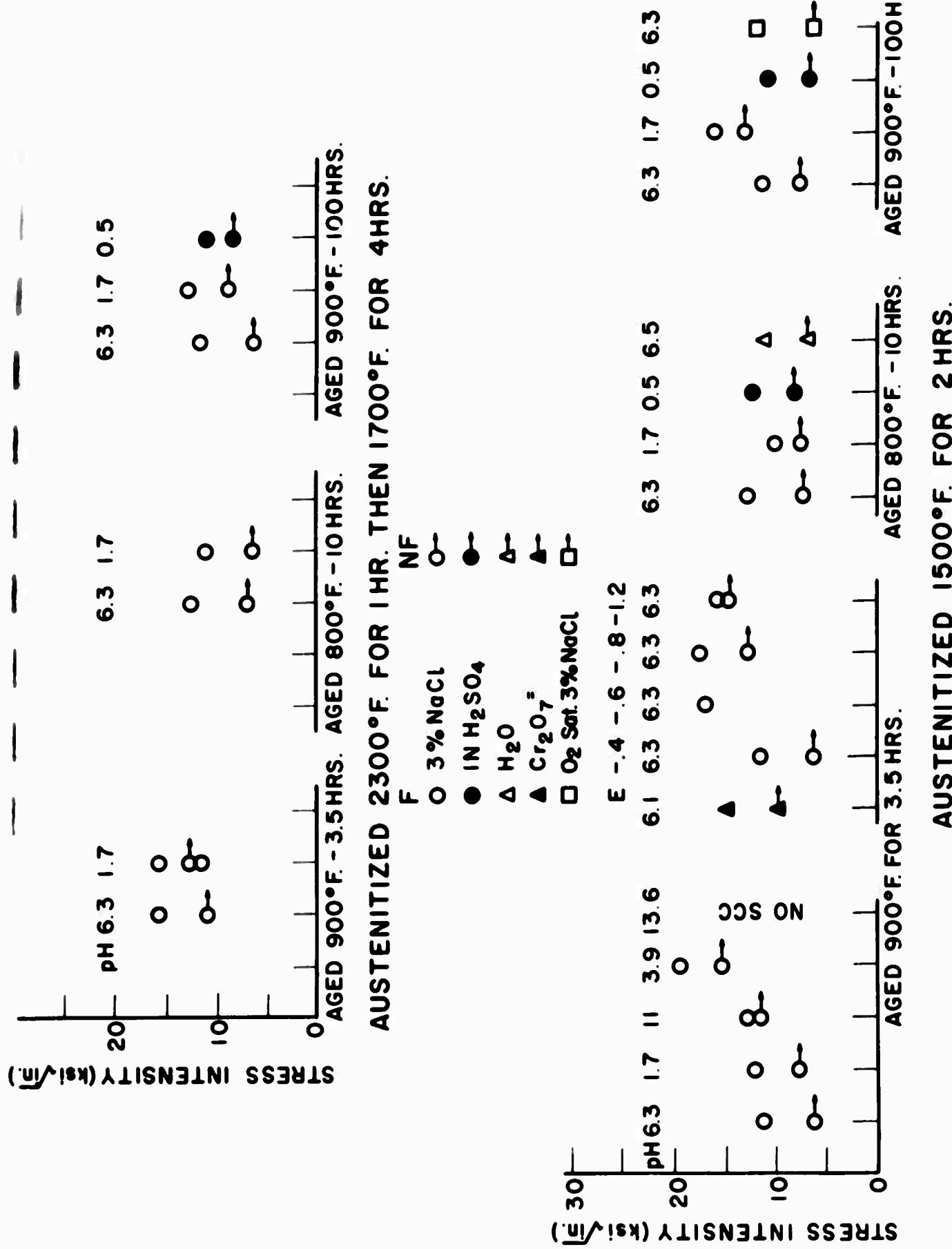


Fig. 10. K_Ic of 18% Ni Maraging Steel in Various Heat Treated Conditions and Environments.